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High Purity Iron.

UNDER ideal conditions, a study of the properties of the binary and other alloys of a metal would be preceded by and based upon an accurate determination of the properties of the pure metal itself; but, as is well known, in the study of the ferrous alloys the properties of pure iron have in practice to be replaced by approximations derived from measurements made on the purest forms of iron available. The standard of purity with which investigators have had to be satisfied has shown a steady advance over a period of years. Many years ago there was a tendency to regard iron produced electrolytically as being necessarily of high purity, but later it was found that unless extreme precautions were taken, with consequent limitation of the amount of metal produced, many forms of commercial iron would rival electrolytic iron for consideration as typically "pure iron." Later, claims have been made for the extraordinary purity of carbonyl iron, but such analyses as are available suggest that this product rarely, if ever, contains more than 99·90 per cent. of iron.

As each type of iron in turn faces improved methods of critical examination its deficiencies are revealed. The pursuit, first, of the 0·9 in the figure 99·9 per cent. and then of the additional 0·09 which has not yet been reached, is admirably described in the recently published book "The Metal-Iron," by H. E. Cleaves and J. G. Thompson. These authors also summarise the properties of the purest irons on record, emphasising the necessity for actual measurements on high purity iron, rather than extrapolation of the results obtained on a series of impure specimens. The recently produced aluminium, zinc, and lead of high purity are more like new metals than mere improvements on existing products, and similarly our knowledge of the effect of small amounts of impurities on iron is not sufficient to enable us to forecast with any certainty many of the properties of the pure metal.

Researches at the National Physical Laboratory have for many years been directed towards the production of iron of the highest possible degree of purity, and an account of the work has recently been published in the "Proceedings" of the Royal Society by Dr. Adcock and Mr. Bristow. All those who have collaborated (and the authors make personal acknowledgment of the part played by about twenty-five past or present members of the N.P.L. staff) are to be congratulated on the outcome of their work. The latest product, "batch No. 5," of high purity iron was examined by chemical and spectroscopic methods for thirty-one separate elements (including gaseous impurities), of which about fifteen were present at least in very small traces, the eight of these occurring in amounts sufficient to be estimated chemically, making up a total of 0·0113 per cent. This iron, containing over 99·98 per cent. of the metal, is probably the purest that has ever been made. It will be understood that when iron of this purity is under investigation the problem of how to keep it free from contamination during the preliminary treatment necessary to bring it into a suitable form and condition for the tests required becomes a matter of very great difficulty. In this connection it has been necessary to develop special refractories and an elaborate technique of vacuum melting and solidification. The high purity iron was remarkably soft and ductile and after preliminary hot rolling could be cold rolled into foil. After 99 per cent. reduction in thickness by cold rolling without annealing there was no indication that a limit had been reached. An abstract of some of the principal physical properties of this high purity iron is appended for convenient reference. Some of the measurements reported raise problems which are as difficult as those which the research set out to solve. The condition of the material used for tensile testing is not stated, but the difference in tensile strength between the two specimens reported was probably due to difference in grain size and crystal orientation. The alpha-gamma transformation, measured with all precautions, was found to occur, not at a constant temperature, but over a definite temperature range. This apparently simple phase change appeared, therefore, to be a more complex transformation, the nature of which will require further investigation. Measurements of electrical resistivity showed that the highest conductivity is not necessarily associated with the purest form of an element. The most interesting points arose, however, in connection with measurements of the magnetic properties. It may be recalled that, some years ago, Cioffi obtained material of very high maximum permeability by prolonged treatment of high purity iron in moist hydrogen at 1475 deg. Cent. It was not known how the hydrogen treatment produced this remarkable result. The treatment improved the purity, increased grain size, induced mechanical softness, and, of course, saturated the specimen with hydrogen. Yensen attributed the high permeability and low hysteresis loss of the hydrogen-treated iron to its improved purity. Cioffi held the view that absorbed hydrogen was mainly responsible. In 1934 he obtained specimens of hydrogen-treated iron having a maximum permeability of 280,000 and for various reasons tended to give more weight to the idea that the purity of the iron was responsible for the high figure. Now, the very careful work of the National Physical Laboratory has failed to obtain in iron of high purity a value of more than about one-tenth of this, and the former explanation, that the high value is due to absorbed hydrogen, may again be said to hold the field. The opinion which has been expressed that the maximum

permeability of a single crystal of absolutely pure iron would be 500,000 now seems unlikely to be correct.

The authors, however, have wisely refrained from speculative explanations in connection with their observations. Such discussions could more suitably be taken up in separate communications, and it may confidently be expected that an adequate supply of high purity iron of the standard now attained will give rise to fresh work in many directions.

Specimens supplied to the Technical High School at Aachen have already enabled Cornelius and Esser to report in the January number of *Archiv für das Eisenhüttenwesen* that the "abnormal" structure observed in carbonyl-iron steels is not due to nitrogen or other impurity derived from the carbonyl iron, but is a property of the high purity iron itself connected, it is supposed, with the very high rate of diffusion of carbon in the pure metal. In steels of such high purity the carbon rapidly migrates towards the first cementite particles to be precipitated, and so causes balling up of the cementite and the occurrence of ferrite even in hypereutectoid compositions.

PHYSICAL PROPERTIES OF IRON OF HIGH PURITY.

The following data have been abstracted from the results of tests reported by F. Adcock and C. A. Bristow in their paper on "Iron of High Purity" ("Proceedings" of the Royal Society, A, Vol. 153, page 172, December, 1935) :—

Iron Batch 5.

Total impurities (31 elements, including O, N, and H) = 0·0113 per cent. + traces.

Density at 19 deg. Cent., 7·871 ± 0·002 g. per c.c.

Coefficient of expansion at 25 deg. Cent., $10{\cdot}8 \times 10^{-6}$.

Temperature range, deg. Cent.	Coefficient of expansion per deg. Cent.
0–100	$11{\cdot}2 \pm 0{\cdot}5 \times 10^{-6}$
0–150	$11{\cdot}8 \pm 0{\cdot}5$
0–200	$12{\cdot}0 \pm 0{\cdot}2$
0–250	$12{\cdot}4 \pm 0{\cdot}2$

Electrical resistivity* at 20 deg. Cent.:

Cold rolled and annealed specimens	9·88 and 9·84 microhm–cm.
Temperature coefficient	$8{\cdot}8 + 0{\cdot}051\ t + 0{\cdot}00006\ t^2$

Thermal conductivity :

At 25 deg. Cent.	0·189	cal./cm. sec. deg. Cent.
At 100 deg. Cent.	0·176	" "
At 150 deg. Cent.	0·167	" "

Critical temperatures :

	Range.	Mean.
Ac_3	924–950	937 deg. Cent.
Ar_3	917–905	911 deg. Cent.

Magnetic properties† :

Remanence, Br	8700 Gauss
Coercivity, Hc	0·271 Oersted

Maximum permeability 14,360 at H = 0·5 Oe.

Hardness :

Cold-worked and annealed two hours at 950 deg. Cent.

Brinell	2 mm./20 kg./30 sec.	49 ± 1
Diamond pyramid	136 deg./10 kg./30 sec.	48 ± 3

Tensile strength :

Limit of proportionality, tons per sq. in.	2·0	2·4
Ultimate stress, tons per sq. in.	11·9	9·1
Elongation, per cent. ($l = 4\sqrt{a}$)	49	36
Reduction of area, per cent.	100	100
Modulus of elasticity, lb. per sq. in.	$29{\cdot}7 \times 10^6$	$28{\cdot}0 \times 10^6$

* Other batches of high purity iron gave results varying from 9·40 to 9·90, indicating that the purest has not necessarily the lowest electrical resistance.

† Other batches gave :—

Br	8440 to 8800
Hc	0·115 to 0·183

Maximum permeability 26,500 to 31,400 at H = 0·3, indicating that the purest iron does not necessarily possess the highest permeability.

Spot-Welding.

THE possible importance of spot-welding in automobile and aircraft construction has led to a considerable amount of metallurgical investigation of the subject. The success or otherwise of the method must in the long run necessarily depend on the provision of suitable plant, operating as far as possible in a manner which will secure the most consistent and satisfactory results without too much dependence on the skill and judgment of the operator. The speed at which spot-welding has to be carried out (itself an advantage of the method) makes it essential that the operation should be automatic when once the time of welding, current density, and pressure between the electrodes have been selected.

These factors have received study, in relation to the quality of the welds produced, in a paper by E. J. L. Dussourd on "The Mechanical Properties of Spot-welds" (*Revue de Métallurgie*, 1935, Vol. 32, pp. 171 and 210). He used steels containing 0·08, 0·42, and 0·83 per cent. of carbon, a chromium-molybdenum steel with carbon 0·25, chromium 0·62, and molybdenum 0·23 per cent., and an austenitic chromium-nickel steel containing carbon 0·19, chromium 17·14, and nickel 8·84 per cent., and investigated the welds by means of microscopical and X-ray examination and by shear and hardness tests. The conditions employed in welding were as follows :—

Material in the form of plates, 5 mm. thick (and some 2·5 mm.).

Electrodes, 10 mm. diameter.

Pressure between electrodes, 220 kilos. to 840 kilos. (2·78 kilos. to 10·6 kilos. per square millimetre).

Time of welding, 0·01 to 0·06 minute, except for the austenitic steel, for which the times were shorter, but were not measured.

Cooling after welding, under pressure or with pressure released.

In the paper, three-dimensional curves are given in which the shear strength of the weld is plotted against the total energy input in kVA, and the pressure between the electrodes for different times of welding. If R is the total shear strength of the weld and A its cross section measured after shear, while E is the cross section of the electrode :—

(1) The values of R/A and R/E were found to pass through a maximum which was about 10 per cent. higher when the weld was cooled under pressure than when the pressure was released before cooling.

(2) The maximum for R/E was reached after a longer time of welding than that for R/A. This indicates that the increase in strength of the weld is due to increase in cross section of the spot, the strength per unit area of spot having already passed its maximum.

(3) Increase in pressure beyond about 4 kilos. per square millimetre retards effective welding, the explanation put forward being that the electrical resistance across the junction of the plates decreases as the pressure increases.

(4) The best welding effect was obtained when the cross section of the spot-weld was nearly the same as that of the electrode.

Apart from these conclusions and the relevant data which may be applicable to the design of spot-welding machines, the paper brings out several interesting metallurgical features of spot-welds.

Spot-welding constitutes a special case of autogenous welding. It is necessary that fusion should be attained in order that the strength of a spot-weld shall be sufficient. Dussourd shows that welding may take place without fusion. A micrograph in his paper shows that the plane of separation of a weld between two plates of mild steel has completely disappeared without any signs of fusion, but, on testing, the strength of the weld so formed was found to be only about half that of the usual fusion spot-weld in the same material. H. Esser,* in a study of the press-welding of iron and steel, showed that a good weld could be obtained without fusion. In the case of iron cylinders pressed together *in vacuo*, the strength of the weld increased continuously as the temperature rose, except at the temperature of A_3, where there was a sudden drop followed by a further increase. The quality of the weld depended on the character of the surfaces and the grain size; the finer the grain size the better the weld. The weld also improved in strength as the pressure was increased up to 5 kilos. per square millimetre. Dussourd points out, however, that the conditions obtaining in spot-welding cannot be compared with those in Esser's experiments, since the surfaces have not the necessary degree of smooth finish and the pressure per square millimetre at the point of contact of the plates is not equal to the pressure per square millimetre of the electrode on the surface; it is less, to a degree which varies with the thickness. Without heating the metal of the spot-weld to fusion it would, moreover, be extremely difficult to arrange conditions to ensure regular mechanical working of the welding plant.

Thus fusion of the metal of the spot-weld is essential. In cooling, the fused metal contracts within the solid envelope of steel surrounding it, and in consequence there always exists a central cavity, the volume of which should be negligible in comparison with that of the spot-weld. For good conditions of welding Dussourd suggests that the cavity should not exceed 1 mm. in diameter in a spot 11 mm. in diameter. The position of the cavity is governed by the orientation of the work. It is always on the vertical axis of the spot-weld, and in a small spot it may be at the exact centre, but in large spots it tends to rise as high as possible above the horizontal axis of the fused zone. Retention of pressure after the current is cut off tends to diminish the cavity. If the time of welding or current density is increased the cavity increases in size, and in extreme cases the internal pressure of the fused metal may become so great as to burst through the envelope, giving a shower of sparks and leaving a serious defect in the weld.

The pressure exerted by the electrodes affects the electrical resistance of the junction between the plates to be welded, but it also has the effect of supporting the malleable zone surrounding the fused metal. It should, therefore, according to Dussourd, be as great as possible, particularly at the end of the welding operation. It is desirable that the pressure should be kept on after the welding current is cut off.

The difficulties encountered in electric welding of medium carbon steels are even more apparent in spot-welding. The steel with carbon 0·42 per cent., and the chromium-molybdenum steel, both showed a martensitic structure and undesirable brittleness at the welds. Steels containing more than 0·3 per cent. of carbon should not, in Dussourd's opinion, be put to use in the condition as welded. It is necessary to carry out a reheating of the weld. Treatment in a furnace cannot be considered as an industrial solution of this difficulty, and the possibility was explored of producing a tempering effect by a second passage of a current in the machine itself. The factors studied were: Rate and manner of cooling after welding, intensity of reheating, and rate of cooling after reheating. The difference in rate of cooling was obtained by maintenance or release of pressure after the current was cut off; only the strength of current was varied, not the time of heating.

Cooling under pressure increased the strength of the welds in mild steel and in chromium-molybdenum steel, but not in medium and high carbon steels. When reheating was carried out, its most beneficial effect was found to be on welds which had been cooled under pressure. After the reheating itself, however, the pressure should be released immediately and cooling allowed to take place freely. In mild steel spot-welds, reheating caused a slight fall in strength. It effected a slight improvement (although it was not indispensable) in chromium-molybdenum steels, but it was essential for medium and high carbon steels, the medium carbon steel showing the greatest improvement as a result of the retreatment.

If spot-welding machines could be equipped with means of providing a lower current for a second heating immediately after the production of the weld and while the steel is still between the electrodes, the operation of spot-welding would, in the opinion of Dussourd, be applicable to a much wider range of steels than it is at present.

* *Archiv für das Eisenhüttenwesen*, 1930–31, 4, 199.

Flakes in Steel Forgings.

The term "flakes" is applied to the more or less circular or elliptical patches sometimes seen on the fracture of test pieces taken from large forgings. In their transverse section flakes have the appearance of fine hair cracks. Their exact cause is still a matter for investigation, for as they occur mainly in large forgings direct experiment on this question becomes difficult and costly. During the period 1914 to 1918, when production, often from inferior raw material, was increased by every possible means, the prevalence of flakes in alloy steel forgings was a source of great trouble and delay. The fact that a recrudescence of the trouble has been experienced in the last few years is not so easily explicable, but may be connected with economic conditions, which have interrupted the established routine of steel works, or may possibly be due mainly to the extension of the use of large alloy steel forgings in new directions. Whatever the cause, exceptional interest in the subject has certainly been aroused and has been revealed by the considerable number of papers published during 1935.

It was therefore appropriate that Messieurs L. Guillet and M. Ballay should write a summarising report on the subject of flakes in steel, to form the basis of a discussion at the International Congress of Mining, Metallurgy and Applied Geology, held in Paris last October.

In their contribution Guillet and Ballay point out that flakes in the usual constructional alloy steels are mainly found in large forgings, though bars of less than 150 mm. diameter obtained from

100 kilos. ingots have been known to show flakes. In a cylindrical forging the cracks are less abundant at the circumference than at the centre, and fewer flakes are found in the region that was the bottom of the original ingot than in that formed from the top or middle. The cracks constituting the flaws are so fine that they are often invisible to the eye until the surface of the steel section has been etched. This has given rise, in some quarters, to the view that it is not the cracks, but merely the tensions which cause them that pre-exist in the forging after cooling, and that additional mechanical stress, as in testing, or a reduction of resistance, as by etching, is required to produce actual cracks. The cracks, however, are revealed not only by etching, but by magnetic detection, or even by merely wetting the surface with a liquid, such as petrol or water, which oozes out of the cracks after the surface has been dried. The redistribution of internal stress caused by cutting the section has also been regarded as the occasion of the cracking, and this is more difficult to disprove as the cracks are too small and too fine for detection by X-rays. On the other hand, a very complete series of experiments, involving interrupted cooling, described by Houdremont and Korschan, has shown that the temperature at which flakes form must be in the neighbourhood of 200 deg. Cent. Whether or not precautions were taken to eliminate residual forging stresses, the occurrence of flakes appeared to depend on the rate of cooling in the region of 200 deg. Cent.

In discussing the causes of flakes, Guillet and Ballay distinguish between the causes of the tensile stresses and the reasons for the localisation of the cracks. The stresses causing flakes have been attributed to :

(1) Irregular cooling, especially the temperature gradient due to rapid cooling.

(2) Residual forging stresses.

(3) Volume changes accompanying physico-chemical transformations.

These hypotheses are not regarded by the authors as entirely satisfactory, especially if the view, strongly held by Houdremont and Korschan is accepted, that the formation of flakes by rapid cooling after reheating is quite exceptional in a forging which has once been safely cooled to atmospheric temperature at a slow rate without inducing flaws. There is not complete agreement on this question of the immunity from flake formation of a steel in all treatments subsequent to the first successful slow cooling operation, nor on the question of whether flaws, once formed, can be welded up during subsequent forging ; but in both instances there is strong affirmative evidence.

(4) Influence of gas. On this hypothesis, gases dissolved in the liquid steel are not entirely liberated during solidification but part remains in solution or combined with the solid metal. During cooling, either because the solubility diminishes with falling temperature or because the compound dissociates, the gases are liberated ; but if they cannot be eliminated by diffusion as fast as they form they may give rise to stresses which exceed the breaking stress of the material. This theory applied to hydrogen has been developed with great skill, both experimentally and theoretically by Bennek, Schenk and Müller, who showed that the formation of cracks at temperatures below 300 deg. Cent. might occur if the rate of cooling is too rapid to permit of the elimination of hydrogen thrown out of solution in that range of temperature.

This theory does not exclude the assignment of a measure of importance to major and to inter-dendritic segregation, and to non-metallic inclusions in the localisation of flakes, since the tensions arising from the evolution of gas will produce fracture at positions of least cohesion. Structural heterogeneity may still have an important influence, since for a given rate of cooling, fissures can most readily be produced in a heterogeneous steel, while in a very homogeneous steel of the same composition the resulting tensile stresses may not attain to the breaking stress of the material at any point. On the other hand, there do not appear to be general grounds for the view that Giolitti has put forward, which makes the absence of non-metallic inclusions all important in the avoidance of flakes. The presence or absence of inclusions is not the only important factor, though impurities may tend to localise the cracks and perhaps to diminish the minimum value of the tensions giving rise to them.

In continuing their summary of the factors which influence the formation of flakes, Guillet and Ballay indicate the relative susceptibility to the defect of steels of different composition. Carbon steels with 1·0 to 1·5 per cent. of manganese, and silico-manganese steels, have some degree of susceptibility and even low-manganese carbon steel (C 0·3 to 0·4, Mn 0·4 per cent.) has been known to give characteristic flakes on rapid cooling. All constructional alloy steels, *e.g.*, nickel, nickel-chromium, and nickel-chromium-molybdenum, as well as 1 to 2 per cent. chromium steels with or without molybdenum, can give flakes. The appearance of flakes in these steels is very characteristic on account of the contrast between the bright grains of the defect, and the dull, fibrous structure of the surrounding fracture. Flakes have never been reported in austenitic steels, and according to certain authors 12 to 13 per cent. chromium steels never show flakes. Baboscin's experience with nickel steel turbine rotors was that the number of forgings showing flakes increased rapidly as the nickel was raised from under 3 to over 3·5 per cent. Increase in the amount of manganese and of chromium, and the addition of molybdenum to a given nickel-chromium composition, also adversely affected susceptibility to flake formation.

Crucible, acid open-hearth and acid high-frequency furnace steels are usually less subject to flakes than steels made in the basic open-hearth or electric furnace. Susceptibility to flake formation increases with size of ingot and diminishes as the casting temperature and rate of cooling are lowered. The mould dressing, according to Maurer and Korschan, plays an important part possibly on account of its hydrogen or methane content. Susceptibility to flake formation also decreases as the amount of work increases. An ingot which has received eight units of work can be cooled without giving flakes at a rate which gives flakes in an ingot of the same steel which has received only three units of work.

Guillet and Ballay accept the evidence which goes to show that a forging which contains flakes may be re-forged to weld up the flakes, and that subsequently no flakes are formed, whatever the rate of cooling adopted, and that a forging which has been cooled slowly and contains no flakes may be reheated and cooled again more rapidly without inducing the defect.

The practicable means of avoiding flakes cannot take much account of composition or methods of manufacture, as these are governed by other considerations. All precautionary measures at present

in use involve some form of slow cooling of the forging with or without soaking at a temperature below Ac_1. Opinion is divided as to the lower limit of temperature down to which slow cooling must be maintained, and also as to the rate of slow cooling to be employed. Guillet and Ballay recommend 200 deg. or 100 deg. Cent. for preference, and give rates varying from 800 deg. to 200 deg. in ten hours for small forgings to 900 deg. to 100 deg. in 100 hours for large forgings.

The most important advance achieved in the study of this subject during the year 1935 has undoubtedly been the establishment by Bennek, Schenck and Müller of a close connection between hydrogen content and susceptibility to flakes. Its chief importance lies in the fact that if hydrogen is the ultimate cause of the trouble, it may be possible to apply a final treatment to the molten steel with the object of eliminating hydrogen and so rendering the steel immune from the defect. At the same meeting of the International Congress, "Some Remarks on the Formation of Flakes in Nickel-chromium-molybdenum Steels" put forward by I. Musatti and A. Reggiori, gave considerable support to the hydrogen theory, besides throwing light on certain disputed characteristics of flakes. These authors carried out several series of comparative experiments, which may be summarised briefly as follows :—

Composition of Steel Used.

	Per cent.
Carbon	0·3 to 0·4
Manganese	0·4 to 0·6
Silicon	0·3
Chromium	0·6 to 0·8
Nickel	2·7 to 3·2
Molybdenum	0·3 to 0·4

Effect of Rate of Cooling After Forging and of a Second Heating and Cooling.

Blooms, 200 mm. square :

(*a*) Cooled in air after rolling	Rich in flakes
(*b*) Cooled slowly to atmospheric temperature	No flakes
(*c*) Cooled as at (*b*), then reheated to 1000 deg. Cent. and cooled in air	No flakes
(*d*) Treatment (*c*) repeated several times	No flakes

In a similar steel which gave flakes when air cooled after rolling, flakes were also present in blooms (150 mm. square), which had been cooled to 280 deg., reheated to 600 deg., held two hours, and either cooled in air or cooled slowly to 300 deg. Cent., and then in air. Only blooms slowly cooled to ordinary temperatures, either directly or with intermediate soaking, were found to be free from flakes. When these 150 mm. square blooms containing flakes were forged to 80 mm. square and cooled in air, no flakes were observed.

Effect of Hydrogen Content.

To investigate this, hydrogen was introduced into the steel by prolonged heating in an atmosphere of hydrogen at 1150 deg. Cent. Analyses were not conclusive as to how much hydrogen was taken up, but on the results of previous investigations it was assumed that the solubility of hydrogen in steel was 7–8 c.c. per 100 g. at 1150 deg. Cent., and nil at atmospheric temperature.

(*a*) A bar of the nickel-chromium-molybdenum steel was heated in hydrogen for seven hours at 1150 deg. Cent., and cooled in air ; no flakes. Treatment repeated ; no flakes. Treatment again repeated ; flakes found.

(*b*) A rod of the steel, 70 mm. diameter, heated twenty-eight hours in hydrogen at 1150 deg. Cent., *i.e.*, four periods of seven hours, and cooled in air after each ; rich in flakes.

(*c*) A rod, 70 mm. diameter, cut from the 80 mm. bar referred to at the end of the previous section as being free from flakes, was heated in hydrogen for seven hours at 1150 deg. Cent., and quenched in water ; flakes again present.

(*d*) Similar heating in nitrogen for fifty hours in no case gave rise to flakes.

Effect of Chemical Composition.

Specimens in the form of cylinders, 60 mm. diameter and 60 mm. high, of the following steels :— Carbon (C 0·37 per cent.), nickel-chromium (C 0·29, Ni 2·49, Cr 0·86 per cent.), and nickel-chromium-molybdenum (C 0·32, Ni 2·87, Cr 0·70, Mo 0·51 per cent.), were heated together in hydrogen and subsequently treated as follows :—

(*a*) Reheated in hydrogen for seven hours at 1150 deg. Cent., and cooled in water. All showed flakes ; in carbon steel very small but numerous, in nickel-chromium steel larger and more numerous, in nickel-chromium-molybdenum steel even more pronounced, but none came to the surface.

(*b*) Reheated in nitrogen for seven hours at 1150 deg. Cent., and cooled in water. No flakes or other fissures.

(*c*) Reheated in hydrogen for five hours at 1150 deg. Cent., cooled in the furnace in a current of nitrogen to 700 deg. Cent., then cooled in water. In spite of the reduced temperature of quenching and the loss of some hydrogen by diffusion the steel again showed flakes.

(*d*) Reheated in hydrogen for seven hours at 1000 deg. Cent. and cooled in water. Flakes in nickel-chromium and nickel-chromium-molybdenum steels, but not in carbon steel.

(*e*) Reheating in hydrogen at lower temperatures (up to thirty hours at 800 deg. and 150 hours at 400 deg. Cent.), and cooling in water, gave no flakes.

All these observations and experiments support the idea that hydrogen is the cause of the formation of flakes. The probable temperature of their formation is in the range 300 deg. to 200 deg. Cent., so the pressure exerted by hydrogen thrown out of solution, but not eliminated by diffusion, must reach the breaking stress of the steel at this temperature. There is no sudden change in the steel itself to account for failure in this range of temperature ; but Musatti and Reggiori quote the work of Jacquerot and Gagnebin to show that there is a very rapid diminution in the rate of diffusion of hydrogen through iron and steel at temperatures between 300 deg. and 180 deg. Cent. The temperature range in which elimination of hydrogen is impeded is thus exactly that at which flakes have been found to occur.

The avoidance of flakes at present entails a considerable increase in cost of production. Precautionary measures can be applied, but the fact that in spite of extreme care they are not always successful shows that they are not based on a fundamental knowledge of the origin of the defect. The margin between success and failure is too small. The recent demonstration of the importance of taking hydrogen content into account may help to furnish a more secure basis for the future control of flakes in steel forgings, but until the methods of manufacture can be laid down with certainty, it will be necessary to retain the restrictive conditions of cooling at present found necessary.

PAPERS ON "FLAKES IN STEEL" PUBLISHED DURING 1935.

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A. Scortecci, "Some Observations on Flakes in Steel," *La Metallurgia Italiana*, 1935, 27, 275.
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I. Musatti and A. Reggiori, "Some Remarks on the Formation of Flakes in Nickel-chromium-molybdenum Steels," Congrès International des Mines, de la Métallurgie et de la Géologie appliquée, Paris, 1935; *Revue de Métallurgie*, 1935, 32, 531.

Iron-Carbon-Manganese Alloys.

PROGRESS in the technology of alloy steels entails a growing demand for more accurate knowledge of the constitution of the ternary systems consisting of iron and carbon with an additional alloying element. From the practical point of view, one of the most important of these systems, as well as one of the most difficult to investigate, is that consisting of iron, carbon, and manganese. Under the title "The System Iron, Cementite, Manganese Carbide, Manganese," R. Vogel and W. Döring have recently published a study of the constitution of these alloys.[1]

Of the binary systems concerned, the iron-carbon diagram has, of course, been most actively investigated; but, thanks mainly to the work of Dr. Gayler[2], the manganese-iron diagram is also known with a considerable degree of accuracy. On the other hand, knowledge of the manganese-carbon system, largely based on the early work of Stadeler[3] and of Kido[4] is very incomplete, and the authors' first object has therefore been to establish this diagram on a firmer basis of experimental work.

THE MANGANESE-MANGANESE CARBIDE DIAGRAM.

The metal employed contained 98 per cent. of manganese with the following impurities:—Iron 0·8, silicon 0·5, aluminium 0·3 per cent., and traces of carbon, sulphur, and phosphorus.

This was melted in Pythagoras crucibles and the carbon introduced in the form of carbon rod. It was necessary to reduce the time of melting to a minimum as the crucible was strongly attacked. For alloys containing over 3·5 per cent. of carbon it was found necessary first to prepare Mn_3C in a graphite crucible by heating manganese and carbon for half an hour at 1500 deg. Cent. The carbide was then remelted with the necessary amount of manganese. It proved to be very difficult to produce and to preserve large quantities of manganese carbide, since it readily decomposed with the formation of hydrocarbons. The difficulty of taking cooling curves was increased by the fact that thermo-couple sheaths were strongly attacked by the melt, and that there appeared almost always to be some under-cooling of the specimen which could only be corrected for by taking several cooling curves of the same alloy. Microscopical examination was rendered difficult by the extreme brittleness of the specimens, and by their instability in the presence of moisture. A serviceable finish for micro-examination could be obtained by substituting alcohol for water in the polishing operations, but in spite of all precautions the carbon-rich alloys disintegrated after a few days. The equilibrium diagram is shown in Fig. 1.

In considering the diagram, it must be recalled that the manganese used was of a distinctly lower grade of purity than that employed by Dr. Gayler, which contained over 99·9 per cent. of manganese. The transformation temperatures recorded for the manganese are lower, thus:

	Gayler. Mn > 99·9 per cent.	Vogel and Döring. Mn 98 per cent.
Melting point..	1244 deg. Cent.	1230 deg. Cent.
$\gamma \rightarrow \beta$	1191	1140
$\beta \rightarrow \alpha$	742	740

and other temperatures, and limits of composition of the phases, may be similarly affected.

Vogel and Döring observed a secondary arrest

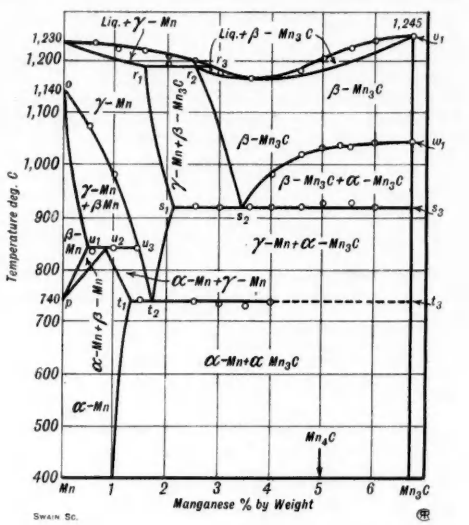


FIG. 1—*The Manganese-Manganese Carbide Diagram.*

on solidification of the 2 per cent. carbon alloy at about 1190 deg. Cent. showing the existence of a gap (which they put at 1·5 to 2·9 per cent. carbon) in the series of solid solutions thought by earlier observers to be a continuous series. They also found that the liquidus dipped to a minimum at 3·5 per cent. of carbon and did not rise to a maximum as Stadeler had supposed. A series of clear arrest points revealed the new fact that manganese carbide undergoes a transformation at 1050 deg. Cent., and these new features of the diagram afford an explanation of the eutectoid arrest at 920 deg. Cent. which had already been noted by Stadeler and by Kido. Microscopic examination placed the lower limit of the eutectoid line $s_1 s_3$ at 2 per cent. carbon. The course of the line $r_2 s_2$ was inferred since, owing to its steepness, there was no thermal arrest, and, owing to the brittleness of the specimens after annealing and quenching, no microscopical examination was possible. The course of the line $w_1 s_2$ was, however, clearly marked by thermal arrests and the eutectoid point was placed at 3·5 per cent. of carbon. At about

[1] R. Vogel and W. Döring, *Arch. Eisenhüttenw.*, 1935–36, 9, 247.
[2] M. L. V. Gayler, *Journ.*, Iron and Steel Inst., 1927 (1) 393; 1933 (2), 293.
[3] A. Stadeler, *Metallurgie*, 1908, 5, 260, 281.
[4] K. Kido, Sci. Rep. Tohoku Univ., 1920, 9, 305.

740 deg. Cent. there was a second horizontal line, not found by former observers, which represents a eutectoid inversion of a γ-manganese solid solution into α-manganese and α-Mn_3C. The temperature of the γ → β change in manganese (which, as already mentioned, was found to be about 1140 deg. Cent.) was lowered by carbon, while that of the β → α change was raised, the two joining at 840 deg. Cent. in the line u_1u_3, which corresponds to the transformation β Mn + γ Mn → α Mn. The heat changes associated with the γ → β and the β + γ → α transformations were clearly marked in the cooling curves, but that of the β → α transformation was very indefinite.

Neither cooling curves nor microstructure gave any evidence of the carbide Mn_4C suggested by Jacobson and Westgren[5] on the basis of X-ray analysis, nor of the carbide Mn_7C_3 more recently proposed by Westgren.

THE SYSTEM FE-Fe_3C-Mn_3C-MN.

This system, previously investigated by Wüst, Goerens and Bain[6], was explored by means of several

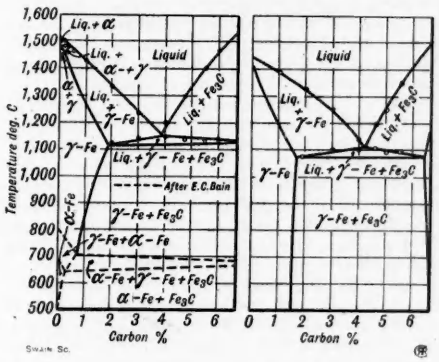


FIG. 2—*The Iron-Carbon-Manganese Diagram. Section with 5 per cent. Manganese.*

FIG. 3—*The Iron-Carbon-Manganese Diagram. Section with 20 per cent. Manganese.*

series of alloys whose compositions, when plotted, passed through the carbon corner of the diagram, or were parallel to the iron-manganese side. There were difficulties in the preparation of the high manganese alloys similar to those already experienced in making the manganese-carbon melts. The iron was therefore first melted in the crucible, the carbon dissolved in it, and finally the manganese added to the melt. The region studied comprised, after solidification, the following four types of crystals :—

(1) Ternary solid solution of carbon and manganese in α (δ)-iron restricted to a small region in the iron corner.

(2) Ternary solid solution of carbon and manganese in γ-iron, covering a very extensive field. This forms the primary crystals of all the usual commercial iron-manganese alloys except ferro-manganese. The pearlite change is suppressed at medium rates of cooling when the manganese exceeds about 5 per cent., so that over a large part of this region of the diagram γ-iron is preserved unchanged to atmospheric temperature.

(3) Ternary solid solution of carbon and iron in γ-manganese, formed only in the manganese-rich region of the diagram (manganese over 70 per cent.).

(4) Solid solution of iron and manganese carbides.

Equilibrium in the Fe_3C-Mn_3C system could not be directly determined on account of the very great

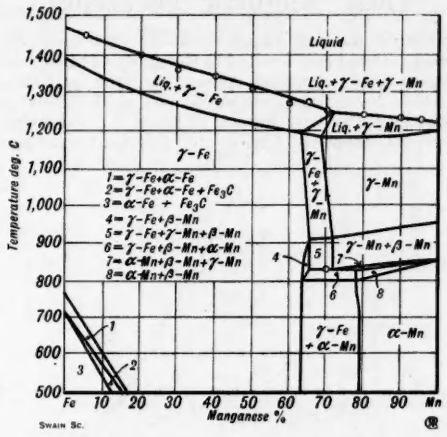

Liquid
Liq. + γ - Fe
Liq. + γ - Fe + γ - Mn
Liq. + γ - Mn
γ-Fe
γ-Mn
1 = γ-Fe + α-Fe
2 = γ-Fe + α-Fe + Fe3C
3 = α-Fe + Fe3C
4 = γ-Fe + β-Mn
5 = γ-Fe + γ-Mn + β-Mn
6 = γ-Fe + β-Mn + α-Mn
7 = α-Mn + β-Mn + γ-Mn
8 = α-Mn + β-Mn
γ-Mn + β-Mn
γ-Fe + α-Mn
α-Mn
Temperature deg. C
Manganese %
Swain Sc.

FIG. 4—*The Iron-Carbon-Manganese Diagram. Section with 1 per cent. of Carbon.*

experimental difficulties. It was assumed that the carbides form a continuous series of solid solutions with a minimum in the freezing point curve.

Sections through the ternary diagram, which include the compositions of all commercial manganese steels, are shown in Figs. 2 to 4. The symbol Fe, of course, refers to the (FeMn) solid solution, and Fe_3C to the isomorphous mixture of carbides

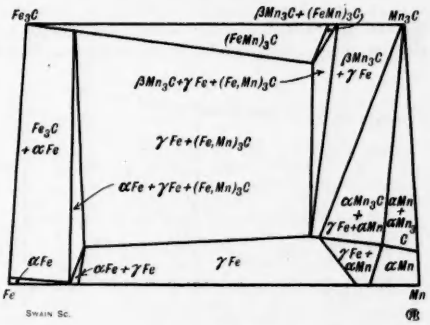


FIG. 5—*Schematic Diagram of the Iron-Carbon-Manganese System at 20 deg. Cent.*

$(FeMn)_3C$. Sections at over 60 per cent. of manganese contain so much detail that they should be consulted in the original paper. Vogel and Döring do not present a quantitative representation of equilibrium conditions at atmospheric temperature, but the schematic diagram, given in Fig. 5, represents the general arrangement of the phase fields. The study of the ternary system is therefore by no means

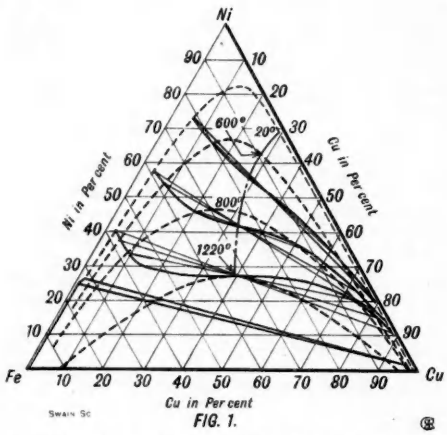
[5] B. Jacobson and A. Westgren, *Z. physik. Chem.*, 1933, B.20, 361 ; A. Westgren, *Jernkont. Ann.*, 1935, 119, 231.

[6] F. Wüst, *Metallurgie*, 1909, 6, 3 ; P. Goerens, *id.* 538 ; E. C. Bain, Am. Inst. Min. Met. Eng., Tech. Pub., No. 467, 1932.

complete, but Vogel and Döring's investigation constitutes a substantial advance on previous knowledge of the constitution of this very important series of alloys.

Nickel-Copper-Iron Alloys and Their Magnetic Properties.

DURING the past twenty years or so considerable attention has been devoted to the study of binary and ternary alloys of nickel, copper, and iron, and many of them have found extensive use in the electrical industry; the best known of these are Permalloy, Mumetal, Isoperm, Invar, and Constantan. More recently a large number of investigations has been made with the object of discovering alloys having improved magnetic properties, and the importance of precipitation-hardening phenomena on these properties has been recognised. In view of the peculiar

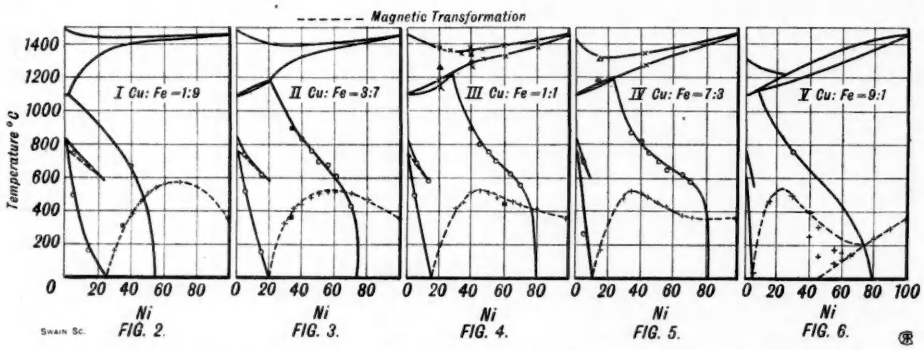


Solubility Isothermals of the Nickel-Copper-Iron System.

magnetic behaviour of nickel-iron alloys it is only natural that attention should be turned to the study of the effect on these alloys of additions of a third metal, and within the past few months several papers have appeared on the equilibria in the nickel-iron-copper system and on the effects of heat treatment on the magnetic properties of these alloys.

The iron-copper system consists of a limited series of solid solutions at each end separated by a large miscibility gap; R. Vogel[1] was the first to point out that this gap gradually closes with addition of nickel, which forms continuous series of solid solutions with both of the other metals. The diagram he put forward for the liquidus and solidus of the ternary system has since been modified only in minor details. But no systematic investigation has until recently been made of the equilibria of these alloys in the solid state. O. Dahl[2] found by X-ray analysis that nickel-iron alloys with 60–85 per cent. nickel have an ordered atomic distribution in the lattice after suitable heat treatment, and later, in collaboration with J. Pfaffenberger[3], he found that the addition of copper up to certain amounts does not destroy this arrangement. O. Dahl, J. Pfaffenberger, and K. Schwartz[4] have just published the results of their investigations on the equilibria in the solid ternary alloys based on measurements of the electrical resistance changes after quenching from high temperatures and annealing at lower temperatures to obtain equilibrium.

Almost simultaneously W. Köster and W. Dannöhl[5] published a complete equilibrium diagram for the system and numerous sectional diagrams illustrating the effects of additions of copper to iron-nickel alloys and of nickel to iron-copper alloys. Although the results of these two groups of investigations differ in some details they are in general agreement on the main points. The miscibility gap at the solidus extends from about 10 to about 95 per cent. copper in a smooth curve through the point in the triangular diagram corresponding to the alloy containing nickel 27, copper 40, iron 33 per cent. With decreasing temperature this curve is displaced towards the nickel corner, reaching 80 per cent. nickel at room temperature—Fig. 1. This also shows the solubility isothermals for 1220, 800, 600, and 20 deg. Cent. (dotted lines) as determined by thermal analysis, resistance measurements, and magnetometric tests. The line crossing them represents the critical points at which two mutually saturated solid solutions merge into a single saturated solid solution. The decomposition of supersaturated solid solutions on cooling is difficult to detect micro-



Effect of Nickel on the Equilibria of Alloys with Constant Iron-Copper Ratios.

scopically even at high powers, the two constituents being almost unidentifiable even after prolonged annealing; the chief difference observed under the microscope is the collecting together of clusters of

fine crystallites similar to that observed in iron-nickel-aluminium alloys, but the decomposition of the solid solution is accompanied by greater ease of etching, a roughening of the surface, and a development of surface colours.

Like the iron-nickel alloys, the ternary alloys can be divided into two groups, the boundary being a straight line joining the 28 per cent. nickel point on the iron-nickel side with the copper corner of the ternary diagram—Fig. 1. Alloys on the iron side of this line are converted from the α to the γ state on heating through temperatures between 900 and 580 deg. Cent. and on cooling revert to the α state at much lower temperatures. Alloys on the nickel side of this line become ferromagnetic on cooling at about the same temperature at which they lose their magnetism on heating, and the temperature-resistance curves show a sharp point of inflection at this temperature, as well as at a higher temperature corresponding with a change from a duplex structure to a homogeneous solid solution. From these curves the sectional equilibrium diagrams shown in Figs. 2–8 have been constructed; the dotted line represents the temperature of magnetic transformation and the other lines the phase boundaries. Figs. 2 to 6 show the effect of nickel additions on the phase boundaries of the system for definite iron-copper ratios of from 9:1 to 1:9. The boundary of the γ solid solution region, which is concave to the composition axis with a high iron-copper ratio, gradually becomes convex thereto as the copper is increased in proportion to the iron. Figs. 7 to 13 show the decrease in the size of the two-phase region and its displacement towards lower temperatures with increase in the nickel percentage.

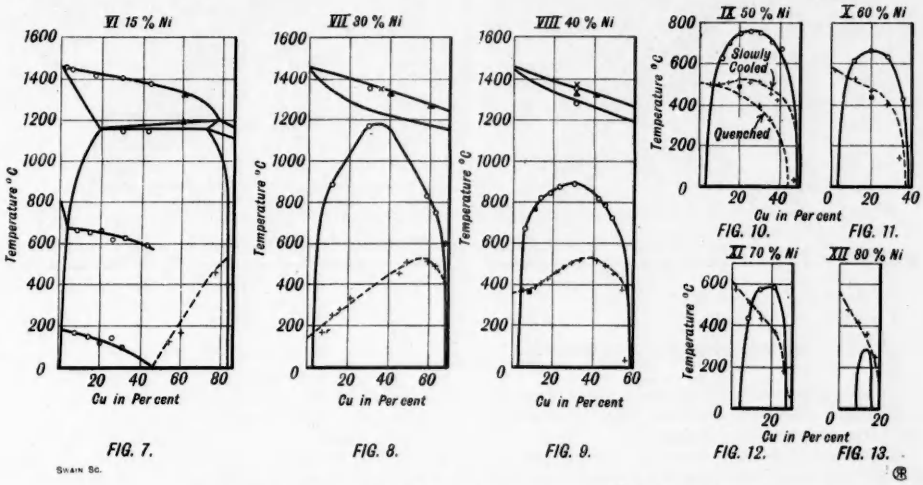
These diagrams clearly indicate the range within which precipitation-hardening should be possible, and tests on these lines by Dahl, Pfaffenberger and Schwartz[4] have confirmed this possibility. When alloys within the two-phase field at room temperature and the homogeneous range at high temperatures are quenched from 1050 deg. Cent. and then annealed at gradually increasing temperatures the resulting temperature-electrical resistance curves show two points of inflection, the first at about 400 deg. Cent. corresponding to the so-called "Permalloy effect," and the second, at 500–700 deg. Cent., to the separation of the second phase from the supersaturated solid solution. The actual increase in hardness in the latter range in no case exceeds about 80 per cent. of the hardness of the quenched alloys; the latter have a Brinell hardness of 70–140, whereas after precipitation hardening a maximum of 200 can be obtained.

Fig. 14 shows the ranges within which the various hardening effects are observed.

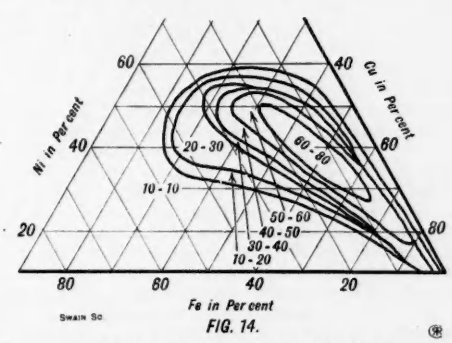
Of far greater practical importance, however, are the changes in the magnetic properties which can be produced by heat treatment of alloys within the duplex range. O. von Auwers and H. Neumann[6] have studied the effects of the following two heat treatment procedures on the initial permeability (μ_o), coercivity (H_c), and induction (B) of alloys with a relatively high nickel content:—

(I) The Permalloy treatment, consisting of annealing at 900 deg. Cent. for one hour, cooling to 625 deg. Cent. and quenching.

(II) A high temperature treatment, consisting of annealing for two hours at 1100 deg. Cent. and furnace cooling.



Effect of Copper on Nickel-Iron Alloys of Constant Nickel Content.



Increase in Hardness Produced by Precipitation Treatment.

When the value of μ_o obtained after either treatment is plotted against the composition there is a definite copper content for every nickel content at which a very large and sudden increase in μ_o is obtained ; with more or less copper than this optimum amount μ_o has very much smaller values. In the

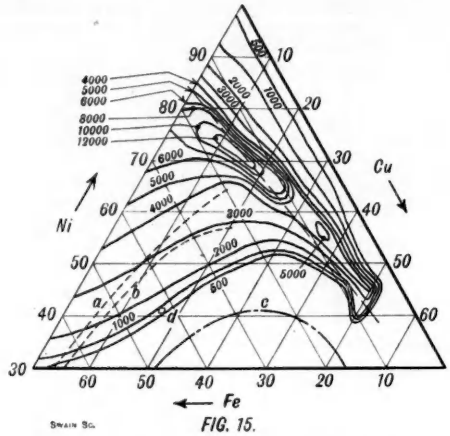


FIG. 15.

Initial Permeability (Heat Treatment I).

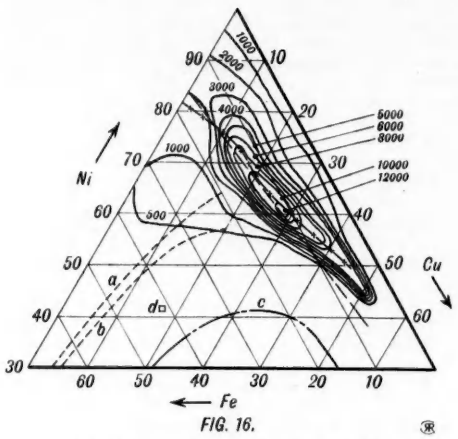


FIG. 16.

Initial Permeability (Heat Treatment II).

Table I are given the maximum values of μ_o for various nickel contents and the copper contents necessary to produce this maximum.

Lines of equal μ_o are reproduced in Figs. 15 and 16, from which and from Table I it will be seen that there is a long narrow range of composition, extending with heat treatment I from 78·5 per cent. nickel and with heat treatment II from 88 per cent. nickel along the copper side to about 50 per cent. copper, 5 per cent. iron, within which μ_o has optimum values, the actual maximum being obtained with 70 per cent. nickel for heat treatment I and with 60–65 per cent. nickel with heat treatment II.

From Figs. 17 and 18, showing lines of equal coercivity, it is apparent that minimum H_c occurs in alloys with less than 70 per cent. nickel at the same compositions as maximum μ_o. After heat treatment I H_c reaches its minimum value at 70 per cent. nickel and after heat treatment II at 45 per cent. nickel. Below 70 per cent. nickel the line of zero magnetostriction coincides with maximum μ_o and minimum H_c, but above 70 per cent. nickel it bends towards the 82 : 18 nickel-iron alloy.

Dahl, Pfaffenberger, and Schwartz[4] have more closely studied the magnetic properties of alloys with a nickel content of 35–55 per cent. Within this range alloys containing more than 6 per cent. copper have a permeability which is independent of the field strength, *i.e.*, they exhibit the characteristic " Isoperm " behaviour. Certain alloys within this range have a further characteristic magnetic property, viz., an extraordinarily high coercivity after suitable heat treatment. If a comparison is made of the coercivity values obtained after air cooling and furnace cooling respectively from 1050 deg. Cent., the extraordinary effect of the latter treatment on alloys with

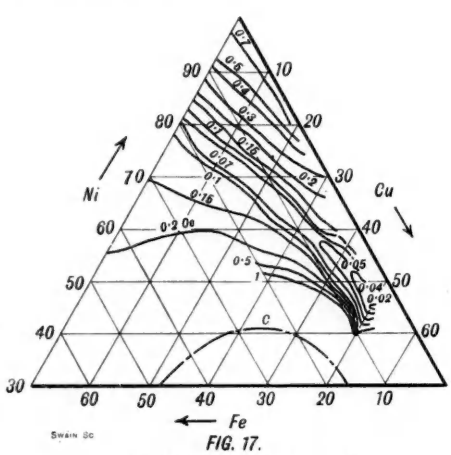


FIG. 17.

Coercivity (Heat Treatment I).

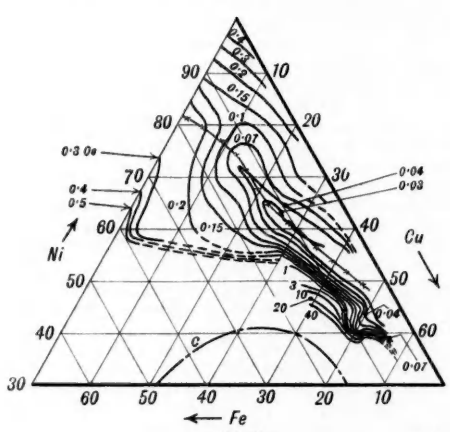


FIG. 18.

Coercivity (Heat Treatment II).

about 60–65 per cent. copper and 10 per cent. iron is clearly indicated by values of H_c of 300–400 Oersted, compared with values of about 0·05 Oersted for alloys with 20–30 per cent. copper and 10 per cent. iron.

In Table II the magnetic properties of some of these alloys are compared with those of various steels.

TABLE I.—*Effect of Composition on Permeability of Nickel-copper-iron Alloys.*

Nickel, per cent.	Heat treatment I.		Heat treatment II.	
	Copper, per cent.	μ_o.	Copper content.	μ_o.
80	4	8,000	9	3,500
76	6	11,000	12	5,000
70	18	13,500	18	9,000
65	23	9,500	22	11,800
60	30	4,500	29	12,000
55	38	5,200	38	6,200
50	42	5,000	45	4,000
45	50	5,000	50	4,200

TABLE II.—*Magnetic Properties of Nickel-copper-iron Alloys Compared with those of Various Steels.*

Copper, per cent.	Nickel, per cent.	Iron, per cent.	Remanence B_r.	Coercivity H_c.	$B_r \times H_c$.
55	37	8	1,650	90	149×10^3
52·5	35	12·5	2,500	215	538
48	32	20	3,500	185	645
30	20	50	4,000	20	80
45	40	15	3,450	172	593
40	40	20	4,800	155	745
60	30	10	1,430	410	586
65	25	10	1,750	360	630
Carbon steel			7,750	52	450×10^3
Tungsten steel			10,500	62	650
Chromium steel			10,300	59	580
Low cobalt steel			9,400	92	880
High cobalt steel			8,500	185	1500

It will be seen that the best alloys have values of $B_r \times H_c$ which are equal to those of tungsten and low cobalt steels; the high coercivity overcomes the low remanence if the magnets are kept relatively short to avoid demagnetisation. An advantage of the copper-nickel-iron alloys over the magnet steels is their ready workability both before and after hardening, a very important point in many cases. A still further advantage is that both the coercivity and remanence are improved by cold rolling; thus when a 30 per cent. extension is given to the 40 : 40 : 20 copper-nickel-iron alloy, after hardening by slow cooling from 1050 deg. Cent., H_c is increased from 155 to 177, B_r from 4800 to 5600, and $B_r \times H_c$ from 745 to 992×10^3.

Good magnetic properties can also be obtained in these alloys by air cooling from 1050 deg. Cent. and annealing at temperatures between 500 and 700 deg. Cent. for two hours, the necessary temperature depending on the composition. The maximum Brinell hardness is obtained by this treatment when the amount of precipitated constituent is a maximum, *i.e.*, at about 600 deg. Cent., but the maximum magnetic hardness is obtained at higher temperatures (700 deg. Cent.), *i.e.*, when re-dissolution has already begun. The effect of particle size of the precipitated constituent on the electrical and magnetic properties of these alloys is remarkable. Thus for the 30 : 60 : 10 nickel-copper-iron alloy the resistance after slow cooling in the furnace from 1050 deg. Cent. is only half that of the quenched alloy, whereas the coercivity is 315 Oersted in the first case and 0·15 Oersted in the second. After air cooling the resistance is 17 per cent. less than that in the slowly cooled condition and the coercivity only 0·2 Oersted. On annealing at 650 deg. Cent. the resistance falls to a minimum in one hour and the coercivity increases to 190 Oersted; after forty hours practically no further change occurs in the resistance, but the coercivity rises still further to 380 Oersted.

According to H. Neumann[7], a still further improvement in the magnetic properties may be effected by annealing the cast alloys at 1000 deg. Cent., quenching in oil, and reheating at 600 deg. Cent. for two hours. Thus after this treatment on the 20 : 68 : 12 nickel copper-iron alloy H_c is 462, B_r 1920, and $B_r \times H$ 897×10^3, and on the 20 : 62 : 18 alloy H_c is 445, B_r 3020, and $B_r \times H_c$ 1343×10^3.

Further work is being carried out along these lines and is stated already to have led to interesting results, which are shortly to be published.

REFERENCES.

1 *Z. anorg. Chem.*, 1910, 67, 1.
2 *Z. Metallkunde*, 1932, 24, 107.
3 *Ibid.*, 1933, 25, 241; *Metallwirtschaft*, 1934, 13, 527, 543, 559.
4 *Metallwirtschaft*, 1935, 14, 665.
5 *Z. Metallkunde*, 1935, 27, 220.
6 *Wiss. Veröff, Siemens-Werken*, 1935, 14 (2), 93.
7 *Metallwirtschaft*, 1935, 14, 778.

Fissures in Railroad Rails.

A PROGRESS report by Professor Herbert F. Moore has recently been issued on the Joint Investigation of Fissures in Railroad Rails conducted by the Engineering Experiment Station, University of Illinois, in co-operation with the Association of American Railroads and the Rail Manufacturers' Technical Committee.*

Considering the large number of rail failures due to transverse fissures which occur annually on railroads in the United States of America, many of these failures having led to serious accidents and loss of life, the publication of this report must be considered of very great importance. It should serve to arouse both railway officials and steel manufacturers to a better realisation of the fact that safeguards against the primary cause of these failures have now been available for some years. One of the great stumbling blocks in the way of this realisation has been the persistence of the theory, which was advanced by early investigators of such failures, that a transverse fissure could start from the interior of the head of a sound rail under the action of wheel loads only, without any pre-existing crack or defect in the metal to act as a starting point.

The results of the experiments described in this report now lead to the conclusion: "At the present time, and subject to revision of belief by final results from the test rails in service, it is felt that the rails investigation has yielded fairly convincing evidence that fissures in rails usually originate at shatter cracks which form in the rails during the process of manufacture."

The existence of internal fissures or shatter cracks in certain rails as delivered from the mill (such defects being similar in nature to the "snow flakes" occasionally met with in alloy steel forgings) has now been known for many years, but this report appears to be the first authoritative statement based on

* Published by the University of Illinois, Urbana, Illinois, U.S.A. Pp. 26. Price 15 c.

definite experimental evidence as to a shatter crack being the necessary origin of a transverse fissure failure.

The report opens with a reference to some previous work by Hoersch on a mathematical analysis of the stresses set up in a rail under rolling wheel loads, and to the view that shearing stresses are probably the main cause of the development of a transverse fissure from a shatter crack.

The main sections into which the experimental investigation of the Committee have been divided are described as follows :—

(1) A laboratory study of the mechanism of the formation of internal fissures in rails under repeated wheel loads.

(2) Field studies of the range of magnitude of wheel loads found in service.

(3) A critical study of acceptance tests for rails at the mills, and a study of the problem of a non-destructive test for rails.

(4) Tests of rails made by special processes designed to prevent internal shatter during manufacture.

(5) Service tests under traffic. These are now in progress, and the results after some years may modify some of the conclusions drawn from laboratory tests.

Section (1).—A large number of newly rolled rails were submitted to the Committee by various manufacturers, and 6in. lengths from each rail were examined for the presence of shatter cracks by deep etching of horizontal slices cut from the head. A footnote refers to the question as to whether the defects revealed by deep etching are actual pre-existing cracks. That actual cracks were present before etching was proved by micro examination of unetched slices, but no mention is made of Rawdon's magnetic iron dust method of detection, which definitely distinguishes cracks from other defects revealed by deep etching.

About 25 per cent. of the rails examined were found to contain shatter cracks; but it is pointed out that regular output should not be judged by this figure, since certain of the rails had been manufactured under definitely unfavourable conditions as being especially suitable for study.

A special testing machine was used in which short lengths of rail could be drawn backwards and forwards under a loaded wheel in such a manner as to permit of various combinations of direct and bending loads, and the results of many tests made with the machine are described. Rails which failed under test did so with two distinct types of fracture, *i.e.*:

(*a*) By fatigue fracture starting from the tread, and

(*b*) By a fatigue fracture starting from a nucleus in the interior of the rail head.

(*a*) was typical of rails free from shatter and only occurred under high flexural stresses (above 60,000 lb. per square inch); (*b*) was typical of some of the rails which had been found to contain shatter and occurred under much lower flexural stress and computed shearing stresses of 20,000 lb. to 40,000 lb. per square inch. These latter results are compared with tests of specimens cut from rail steel and subjected to cycles of shear from zero to a maximum, which gave an endurance range of 55,000 lb. to 65,000 lb. per square inch, such an endurance range being approximately halved by stress concentration at the edge of a small crack.

Section (2).—Measurement of actual wheel loads under passing trains in track were made with a De-Forest type of scratch extensometer attached to the flange of a rail in such a manner as in no way to interfere with the normal track construction. The readings could be reduced to wheel loads by formulæ developed by Dr. Talbot. Preliminary results obtained on two railroads, with 152 lb. and 100 lb. rails respectively, showed that a small percentage of wheel loads may rise to over 50,000 lb. in the former and up to 70,000 lb. in the latter; the highest figures being usually due to flat spots on the wheels. From the rolling load tests in the laboratory it was estimated that any wheel load above 40,000 lb. might, if repeated often enough, spread a shatter crack into a transverse fissure.

Section (3).—Various inspection tests for rails are considered and the interesting view advanced that a slow bend test on a 5ft. or 6ft. length of rail would give more useful information regarding quality than the present standard drop test. Neither drop test nor bend test is, however, capable of definitely revealing the presence or absence of shatter cracks, and the only method at present available is the macro examination of sections from the head. Many non-destructive tests, *e.g.*, electrical resistance, distortion of magnetic fields, &c., were studied, but no method has yet been found capable of detecting an internal crack of less than about 0·5 square inch in area.

Section (4).—Rails treated by two special processes for the elimination of shatter were examined and tested by bend tests and in the rolling load machine. The first process, *i.e.*, controlled cooling of the rails from the mill, is not mentioned by name, but is undoubtedly the Sandberg Oven Process, now in operation at the Steelton and Lackawanna works of the Bethlehem Steel Company and at the Chicago works of the Inland Steel Company in the United States of America, and incidentally at practically all rail works in this country. This process avoids the formation of shatter cracks by equalising the temperature of the rail throughout its section just after passing the critical temperature, followed by subsequent slow cooling. Extended experiments and practice both in this country and in the United States of America have proved that its use is a complete preventative of the formation of shatter. Rails supplied to the Committee which had been treated by the process showed no shatter and did not develop fissures under the rolling load tests, while some of the companion rails hot-bed cooled from the same heats of steel showed shatter and developed fissures under the rolling load test.

The second process (referred to in a footnote as the Brunner normalising process) claims to avoid shatter by means of grain refinement, obtained by allowing rails from the mill to cool just below the critical temperature, reheating to just above it, and then cooling normally in air. Among the rails submitted to the Committee treated by the process, a few showed the presence of shatter cracks, but did not develop fissures under the rolling load test, while companion hot-bed cooled rails showed many shatter cracks and developed fissures under the rolling load test.

Section (5).—No particulars are given regarding these tests.

The report concludes with some test results of rail steel at low temperatures and with notes on many miscellaneous investigations which are not reported.

An appendix gives a bibliography on internal fissures in rails.

The Effect of Additions of Nitrogen on the Properties of High Chromium-Iron Alloys.*

By J. A. JONES, M.Sc., F. Inst. P.

FOLLOWING the publication by Adcock in 1926 of the results of investigations into the influence of nitrogen on chromium and some chromium-iron alloys, considerable research has been carried out in America, particularly by Franks on the effect of nitrogen on the properties of commercial cast and wrought iron alloys of high chromium content ("The Effect of Nitrogen on Chromium and Some Chromium-Iron Alloys," F. Adcock, *Journal*, Iron and Steel Institute, 1926, 2, page 117; "Chromium Steels Improved by Nitrogen," Russell Franks, *Iron Age*, September 7th, 1933, page 10; "Chromium Steels of High Nitrogen Content," Russell Franks, American Society for Metals, 1935, Preprint No. 30). Franks has made claims that castings of high chromium content, to which additions of nitrogen have been made, show improved properties, particularly from the viewpoint of grain refinement associated with increased strength and ductility. The resulting castings are said to possess sufficient ductility to withstand rough usage without cracking. This is not always the case with castings of plain chromium steels or irons.

Franks has also claimed that similar improvements are obtained in the properties of wrought alloys of high chromium content as a result of the additions of nitrogen. Again, the benefits extend to the austenitic 18/8 chromium-nickel alloys.

It is generally believed that the nitrogen present in these alloys exists in combination with the chromium as a highly stable nitride, which behaves in many respects similarly to a carbide. The nitrogen is added by means of a ferro-chromium-iron alloy, containing a high percentage of nitrogen.

The results of the later investigations by Franks show that the addition of 0·10 per cent. nitrogen to the ductile low-carbon alloys containing 12 to 14 per cent. chromium has a marked effect in increasing hardenability without seriously decreasing ductility. In the case of the ferritic chromium-iron alloys containing over 16 per cent. chromium, moderate percentages of nitrogen serve to refine the grain structure and ensure appreciable toughness in the alloys. Particular attention is paid by Franks to the effect of additions of nitrogen to the 25 per cent. chromium alloys, and he states that the influence of 0·15 to 0·25 per cent. nitrogen in producing fine grain alloys with improved toughness and resistance to deterioration at elevated temperatures has been demonstrated.

In view of the importance of the results obtained in America, tests were carried out in the Central Research Department of the United Steel Companies, Ltd., on various high chromium-iron alloys containing nitrogen.

The alloys were made as 18 lb. ingots from melts carried out in a 35-kVA high-frequency furnace of the Ajax-Northrup spark-gap type. The nitrogen was added invariably as high nitrogen ferro-chromium in the cold charge and no loss of nitrogen was obtained in melting down. For example, using a ferro-chromium alloy containing 66 per cent. chromium and 0·72 per cent. nitrogen in the ratio of one-third of the total charge, the resulting ingots would always give a nitrogen figure very close to the theoretical yield of 0·24 per cent.

The alloys investigated cover a range of chromium content from 18 to 30 per cent.

18 PER CENT. CHROMIUM IRONS.

It was anticipated that the addition of nitrogen to a ferritic chromium iron would have the effect of rendering the alloy martensitic in a similar manner to the addition of low percentages of nickel. The effect of additions of nitrogen was therefore examined from the point of view of the possibility of producing an improved hardenable high chromium iron.

Two ingots were made containing approximately 18 per cent. chromium, one with no nitrogen additions and the other with high nitrogen ferro-chrome additions, to give approximately 0·20 per cent. nitrogen in the final iron.

Analyses.

	Ingot 756. Per cent.	Ingot 757. Per cent.
Carbon	0·11	0·11
Manganese	0·42	0·43
Silicon	0·35	0·28
Chromium	17·92	17·84
Nitrogen	0·01	0·20

The ingots were preheated to 950 deg. Cent., and after raising to forging temperature forged fairly gently from 1150 deg. to 950 deg. Cent., partly to 1in. and partly to §in. diameter bars.

Although the surfaces of the bars appeared sound after forging, it was found on machining that the bars from the ingot containing nitrogen were so badly cracked that it was impossible to carry out mechanical tests on the material. Sufficient sound sections were obtained, however, to carry out a series of heat treatments on discs ½in. thick, and to determine the Brinell hardness. The results of these tests were as follows:—

Temperature (one hour). Deg. Cent.	Brinell hardness numbers. Ingot 756. Air cooled.	Ingot 756. Oil quenched	Ingot 757. Air cooled.	Ingot 757. Oil quenched
700	146	143	217	212
750	143	143	207	196
800	140	140	192	179
850	137	137	183	170
900	137	143	255	286
950	137	149	387	402
1000	143	153	418	430
1050	174	183	418	418

The addition of 0·2 per cent. nitrogen thus renders 18 per cent. chromium iron hardenable, the maximum hardness being obtained after cooling from a temperature of 1000 deg. Cent.

In view of the presence of cracks in the bars forged from ingot 757, a further ingot was made of similar analysis, but slightly lower in carbon and nitrogen contents:—

Analysis.

	Ingot 830. Per cent.
Carbon	0·09
Manganese	0·44
Silicon	0·29
Chromium	18·08
Nitrogen	0·16

The forging of this ingot was carried out at another Works, with precisely similar results to those obtained with the previous ingot.

The results indicate that the difficulties of mani-

* A communication from the Central Research Department, the United Steel Companies, Ltd., Stocksbridge, near Sheffield.

pulation of nitrogen-bearing 18 per cent. chromium irons would be greater than those met with in 18 per cent. chromium irons containing approximately 2 per cent. of nickel.

22 PER CENT. CHROMIUM IRONS.

The higher chromium irons of low carbon content generally give particularly low notched-bar impact figures and in view of the fact that such alloys remain ferritic over a very wide range of temperature of treatment, little improvement is brought about normally by heat treatment.

The alloys are also susceptible to grain growth and embrittlement when exposed for prolonged periods at high temperatures. Tests were therefore put in hand on the effect of nitrogen additions, with a view to ascertaining whether any improvement in quality would result. Two ingots containing approximately 22 per cent. chromium were made in the same manner as that adopted for the previous ingots, one with no nitrogen addition and the other with an addition of approximately 0·2 per cent. nitrogen. Tests were carried out on heat-treated sections of ⅝in. diameter bars forged from the ingots. No indication of the presence of cracks was found in these bars.

Analyses.

	Ingot 529. Per cent.	Ingot 1279. Per cent.
Carbon	0·09	0·07
Manganese	0·33	0·38
Silicon	0·25	0·35
Chromium	21·47	21·40
Nitrogen	—	0·21

The 22 per cent. chromium iron shows good ductility figures as measured by percentage elongation, but the notched bar impact figures are particularly low. The addition of 0·2 per cent. nitrogen increases appreciably the tensile strength of 22 per cent. chromium iron, particularly when the alloy is cooled rapidly from temperatures over 1000 deg. Cent.

The tensile strength of the alloy containing nitrogen is, however, still fairly low, whilst the ductility, as measured both by percentage elongation and by notched bar impact figures, is considerably higher than that of the corresponding plain chromium iron.

Tests were also carried out on material from two further ingots of 22 per cent. chromium iron to determine the effect of addition of nitrogen on the susceptibility to embrittlement, on exposure for prolonged periods at 1000 deg. Cent.

Analyses.

	Ingot 775. Per cent.	Ingot 1280. Per cent.
Carbon	0·08	0·07
Manganese	0·40	0·42
Silicon	0·30	0·35
Chromium	22·74	22·30
Nitrogen	—	0·21

The ingots were rolled to strip. Ingot 775 gave entirely satisfactory 10/12 gauge strip, but the material from ingot 1280 developed bad edges at 5 to 6 gauge.

Bend tests were carried out on lengths of strip after exposure at 1000 deg. Cent. for periods of time extending to 168 hours.

Temperature, deg. Cent.	Time at temperature, hours.	Angle of bend without fracture. Ingot 775.	Ingot 1280.
		Deg.	Deg.
As rolled	—	180	180
1000	1	180	180
1000	24	53	45
1000	168	0	31

The 22 per cent. chromium iron containing nitrogen suffers marked embrittlement after exposure to a temperature of 1000 deg. Cent. for periods of twenty-four hours and over, but after 168 hours' exposure at 1000 deg. Cent., the nitrogen-bearing iron retains considerably more ductility as measured by a bend test, than the corresponding iron without nitrogen.

28 PER CENT. CHROMIUM IRONS.

Similar tests to those carried out to test the susceptibility of 22 per cent. chromium irons containing 0·20 per cent. nitrogen were carried out on 28 per cent. chromium irons.

Analyses.

	Ingot 1282. Per cent.	Ingot 1281. Per cent.
Carbon	0·20	0·19
Manganese	0·96	0·94
Silicon	0·32	0·56
Chromium	26·92	28·35
Nitrogen	—	0·27

The ingots rolled satisfactorily to 10/12 gauge

Ingot 529.	Max. stress, tons per sq. inch.	Yield point, tons per sq. inch.	Elongation, per cent. on 2in.	Reduction of area, per cent.	Izod impact figures, foot-pounds.	Average.
As forged	38·2	28·7	27·0	52·0	5, 4, 5	5
Annealed 750 deg. Cent.	40·3	32·3	25·0	54·8	3, 2, 3	3
Air cooled, 900 deg. Cent.	36·2	27·6	30·0	64·0	5, 6, 5	5
Air cooled, 1000 deg. Cent.	38·2	26·7	27·0	31·5	6, 4, 7	6
Water quenched, 1000 deg. Cent.	41·4	32·7	19·5	22·0	6, 5, 6	6
Air cooled after two hours at 1100 deg. Cent.	36·5	28·3	30·5	42·0	3, 4, 5	4

Ingot 1279.	Max. stress, tons per sq. inch.	Yield point, tons per sq. inch.	Elongation, per cent. on 2in.	Reduction of area, per cent.	Izod impact figures, foot-pounds.	Average.
As forged	45·6	38·8	21·0	59·2	9, 5, 12	9
Annealed, 750 deg. Cent.	47·5	34·0	25·5	54·8	7, 11, 9	9
Air cooled, 900 deg. Cent.	41·6	29·2	29·5	66·0	10, 18, 31	20
Air cooled, 1000 deg. Cent.	39·6	24·2	30·0	59·2	8, 12, 11	10
Water quenched, 1000 deg. Cent.	44·9	30·6	31·0	44·4	7, 9, 8	8
Air cooled after two hours at 1100 deg. Cent.	45·0	32·2	33·5	49·6	9, 10, 10	10

strip and lengths of strip were exposed to temperatures of 1000 deg. to 1100 deg. Cent. for periods of time extending to sixteen hours. Bend tests were carried out with the following results :—

Temperature, deg. Cent.	Time at temperature, hours.	Angle of bend without fracture.	
		Ingot 1282.	Ingot 1281.
		Deg.	Deg.
As rolled	—	180	180
1000	1	180	180
1000	8	180	180
1100	1	180	180
1100	8	7	180
1100	16	0	115

The addition of nitrogen to 28 per cent. chromium iron has a beneficial effect in reducing the susceptibility to embrittlement as a result of exposure to a temperature of 1100 deg. Cent., the rate of embrittlement at this temperature, in the case of the iron containing nitrogen, being considerably less than that of the corresponding iron without nitrogen. The fact that the iron containing nitrogen ultimately becomes embrittled at 1100 deg. Cent. indicates, however, that the addition of nitrogen cannot be expected to extend appreciably the safe temperature range of application of the 28 per cent. chromium irons.

HIGH CHROMIUM CAST IRONS.

Tensile, bend, and unnotched bar impact tests were carried out on a number of cast bars containing :

	Per cent.
Carbon	1·50 to 1·85
Manganese	0·55 to 0·65
Silicon	1·70 to 2·60
Chromium	29·5 to 31·5
Nitrogen	Nil to 0·30

All bars were annealed at 1000 deg. Cent., and in this condition gave Brinell hardness numbers of 330 to 350.

Although the nitrogen-bearing irons showed markedly finer grains than the irons free from nitrogen, no corresponding improvement in any of the physical properties examined could be observed.

CONCLUSIONS.

The results of this investigation confirm that the addition of about 0·2 per cent. of nitrogen to the ferritic 18 per cent. chromium irons renders the alloys hardenable in a similar manner to the addition of about 2 per cent. of nickel or an increase in carbon content.

The difficulties of manipulation of such hardenable nitrogen-bearing chromium irons are, however, greater, in our somewhat limited experience, than those met with in similar irons containing nickel.

Laboratory tests have demonstrated that nitrogen produces grain refinement in high chromium irons.

As regards mechanical properties, the addition of nitrogen has been found to produce an improvement in notched bar impact figures in the 22 per cent. chromium alloys, whilst susceptibility to embrittlement during exposure at high temperatures is reduced to some extent.

It has, therefore, been demonstrated, as shown earlier by Franks, that the addition of nitrogen to high chromium alloys definitely modifies their properties. As far as we are aware, no data are available in this country at the moment with regard to service tests on high nitrogen chromium irons.

Definite claims are made in the States with regard to the improved service value of high chromium iron castings containing nitrogen, whilst in the course of the discussion on Franks' paper advantages were claimed from the point of view of easier manipulation into seamless tubing of 25 per cent. chromium irons when containing nitrogen. Greater ease of welding, with additional ductility in the weld, is also claimed as a result of additions of nitrogen. It is emphasised, however, that in order to develop the full benefits from additions of nitrogen, the carbon content must be kept as low as possible, whilst preference is expressed for certain chromium contents as being most responsive to improvement.

It thus remains a matter of some conjecture as to how far these alloys will find useful application and development in this country.

The author wishes to thank Dr. T. Swinden, Director of Research, the United Steel Companies, Ltd., for permission to publish these results.

Books and Publications.

Chromium Plating, with Special Reference to its Use in the Automobile Industry. By O. BAUER, H. ARNDT, and W. KRAUSE. Translated from the German by E. W. PARKER. With an Introductory Chapter by A. W. HOTHERSALL. 266 pages. 1935. London: Edward Arnold and Co. 25s.

A TRANSLATION of the original German edition with some condensation and a very considerable addition to the bibliographical references and annotations, forms the central feature of this English edition. It consists essentially of a description and discussion of results of an extensive examination of typical plated parts of German motor cars. The plated components, together with specially prepared samples of the various basis metals in sheet form, also plated under commercial conditions, were examined for thickness of coating, adhesion, and capacity for deformation, hardness and wear resistance, and resistance to variations of temperature; and were also tested for porosity and behaviour under various corrosive conditions. New tests for determining the porosity of coatings on zinc-base die castings and aluminium alloy castings were evolved, and methods of testing the resistance of the coatings to wear and to variations of temperature were devised.

The subject is dealt with from a metallurgist's point of view, and there is little in the book on the actual process of chromium plating. The metallurgical aspect is of such importance that this omission would not be serious were it not for the title of the book. Since it is the properties of the undercoating which chiefly determine the protective value of chromium coatings, the properties of nickel coatings are discussed in relation to the plating process in the introductory chapter. A more appropriate title to the book would thus seem to have been "The Properties of Nickel and Chromium Coatings on Automobile Components."

The outlook of the original authors appears to have been somewhat restricted. One is inclined to be puzzled, for example, by statements which appear in the various chapters attributing a controlling *rôle* to hydrogen gas in determining hardness, stress, adhesion, and porosity. This theory has, of course, been largely disproved or extended by English, French, and American work during recent years. In order to correct the balance of the book in this respect, an introductory chapter has been written by Mr. A. W. Hothersall. This chapter not only provides a critical summary of the German authors' work, but also constitutes a brief but comprehensive statement of the present position of knowledge on the properties

of nickel, and to some extent also of chromium coatings. It is, in itself, a valuable contribution. Furthermore, the translator's additional notes and references help to give a more complete and accurate picture of the present state of knowledge of the subject than could be gained from the original German text.

The chapter on adhesion, although written originally with an incomplete appreciation of published work, is a useful and novel contribution to a subject on which there is much misunderstanding. Although the critical examination of various methods employed for testing the adhesion of thin coatings did not result in the evolution of a satisfactory method, it led up to the enunciation of an important distinction between degree of adhesion and tendency to peeling, and also to an analysis of the factors which contribute to the capacity of a plated coating to withstand deformation without peeling.

The chapters on porosity and corrosion resistance are of special importance, and it is interesting to note the agreement between the results of these authors and the recent American investigations of Blum and his colleagues in showing a clear relationship between the thickness of commercially produced nickel coatings and their porosity and also between the results of porosity and corrosion tests. It is doubtful if so complete a record of the behaviour under corrosive conditions of platings on steel, brass, bronze, and zinc-base and aluminium castings is available elsewhere in the literature.

The book is printed entirely on art paper, and this probably accounts for the excellent reproduction of the illustrations. The sections of various coatings showing their thickness are exceptionally well prepared and illustrated. Numerous photographs show the results of tests of adhesion, porosity, and corrosion resistance, and provide a very clear picture of the work. Both author and subject indexes are provided, the author index relating to both the text and the bibliography.

The bibliography compiled by the translator is a particularly valuable feature of the book. Altogether, 582 references are collected in chronological order of years, those in any one year being arranged alphabetically under the authors' names. Only specially selected references are included prior to 1920. To those interested in the electro-deposition of nickel and chromium and in the properties of electro-deposited coatings, this bibliography will be of the highest value.

Oxidation and Scaling of Heated Solid Metals. Department of Scientific and Industrial Research. 8vo. Pp. vii+104. 1935. London: H.M. Stationery Office. 2s. 6d. net.

THE study of the oxidation and scaling of metals has of necessity become increasingly important under the economic conditions which have developed. The waste due to metal loss and the labour involved in removing the scale constitute a serious cost to the metal industry.

Much work has already been done in studying the mechanism and rate of oxidation of different metals and alloys in different atmospheres and the nature and properties of the oxide layers formed, as well as methods of preventing oxidation and of removal of scale. With this in mind the Metallurgy Research Board and the Department of Scientific and Industrial Research, acting on the advice of investigators who have had occasion to deal with these phenomena, compiled a critical review of previous work with a view to forming a basis for further investigations.

Six sections of the review deal respectively with the theoretical aspects of oxidation, the constitution and formation of scale on ferrous alloys, the quantitative aspects of the oxidation and scaling of iron and steel at elevated temperatures, and the industrial aspects of the subject.

Dr. U. R. Evans has contributed the introductory section, dealing with surface films and the mechanism of diffusion through these films. Time oxidation curves and the stimulation of oxidation by atmospheric impurities are also discussed in this section. Dr. L. B. Pfeil and Mr. A. B. Winterbottom, in the section on the constitution and formation of scale, show clearly that much further work is necessary on the scaling of alloy steels and the scaling of iron in complex atmospheres. Dr. J. C. Hudson and Mr. T. E. Rooney survey the major researches on the oxidation of ordinary iron and steel in the section on the quantitative aspects of oxidation and scaling. The section devoted to a review of the work done on the oxidation of non-ferrous metals has been contributed by Dr. J. S. Dunn and Mr. F. J. Wilkins. The practical aspects of the oxidation and scaling of ferrous metals and non-ferrous metals are dealt with respectively by Mr. S. A. Main and Dr. L. B. Pfeil, the latter being concerned mainly with the possible metals for heat-resisting purposes. Reference is also made to bright annealing.

Naturally the references form an extremely valuable and essential section of the book, but its value could have been still further enhanced by the inclusion of an index.

It is to be hoped that the Review will be the first of a series of progress reports on the subject. As such it admirably fulfils its purpose.

It is fitting to mention here that when this Review appeared there was already in course of publication an Institution of Gas Engineers Research Fellowship Report (79 pp.) on "The Scaling of Mild Steel in Sulphur-free and Sulphur-containing Furnace Atmospheres," by Dr. H. C. Millett and Professor J. W. Cobb. Some results of this research were made available in advance of publication so that the work does not pass unnoticed in the D.S.I.R. Review, but the full Report deserves the close attention of all concerned with the use of gas-fired furnaces in metallurgical operations such as annealing and other heat treatment.

Mémoires de Métallurgie presentés au Congrès International des Mines, de la Métallurgie et de la Géologie Appliquée. 1935. Pp. 476, 4vo. *La Revue de Métallurgie*, 5, Cité Pigalle, Paris, 9. 75f.

THE first volume of the "Proceedings" of the Metallurgical Section of the recent International Congress, to the success of which reference was made in the December number of THE METALLURGIST, has been issued. A second volume is expected to appear in April next at a pre-publication price of 40f. The completed publication will form a valuable record of the 121 separate communications presented to the Metallurgical Section of the Congress. The present volume contains papers (originating from many countries, but all in French) on the manufacture and properties of iron and steel, of non-ferrous metals, and of alloys both ferrous and non-ferrous, and includes sections on treatment, uses, and tests. Many of these papers contain admirable summaries of the present position of certain metallurgical problems of current interest suggested by the Organising Committee, and collected together in this way they form a valuable record of the many activities of the Congress.